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STUDY OF RESIDUAL BACKGROUND CARRIERS IN MIDINFRARED InAs/GaSb SUPERLATTICES FOR UNCOOLED DETECTOR OPERATION (POSTPRINT)

H.J. Haugan, S. Elhamri, F. Szmulowicz, B. Ullrich, G.J. Brown, and W.C. Mitchel

Electronic and Optical Materials Branch
Survivability and Sensor Materials Division

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6. AUTHOR(S)
H.J. Haugan, G.J. Brown, F. Szmulowicz, and W.C. Mitchel (AFRL/RXPS)
S. Elhamri (University of Dayton)
B. Ullrich (Bowling Green State University)

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Nonstructural Materials Branch (AFRL/RXPS)
Nonmetallic Materials Division
Materials and Manufacturing Directorate
Wright-Patterson Air Force Base, OH 45433-7750
Air Force Materiel Command, United States Air Force

University of Dayton
Department of Physics
Dayton, OH 45469

Bowling Green State University
Department of Physics and Astronomy
Bowling Green, OH 43403

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14. ABSTRACT
The midinfrared 7 ML InAs/8 ML GaSb superlattices (SLs) were grown by molecular beam epitaxy at growth temperatures between 370 and 430 °C in order to study the intrinsic characteristic of background carriers. Grown SLs were all residual *p* type with carrier densities in the low 10^{11} cm^{-2}, and a minimum density of 1.8×10^{11} cm^{-2} was obtained from the SL grown at 400 °C. With increasing growth temperature, the in-plane carrier mobility decreased from 8740 to 1400 cm^2/V s due to increased interfacial roughness, while the photoluminescence intensity increased sixfold due to a decrease in the nonradiative defect densities.

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Study of residual background carriers in midinfrared InAs/GaSb superlattices for uncooled detector operation

H. J. Haugan,[a] S. Elhamri,[b] F. Szmulowicz, B. Ullrich,[c] G. J. Brown, and W. C. Mitchel
Air Force Research Laboratory, Materials and Manufacturing Directorate, Wright-Patterson Air Force Base, Ohio 45433, USA

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The midinfrared 7 ML InAs/8 ML GaSb superlattices (SLs) were grown by molecular beam epitaxy at growth temperatures between 370 and 430 °C in order to study the intrinsic characteristic of background carriers. Grown SLs were all residual *p* type with carrier densities in the low 10^{11} cm^{-2}, and a minimum density of 1.8×10^{11} cm^{-2} was obtained from the SL grown at 400 °C. With increasing growth temperature, the in-plane carrier mobility decreased from 8740 to 1400 cm^2/V s due to increased interfacial roughness, while the photoluminescence intensity increased sixfold due to a decrease in the nonradiative defect densities. © *2008 American Institute of Physics*.
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The development of advanced midinfrared detectors that can operate at room temperature has gained much attention recently due to increasing demand in thermal imaging, pollutant detection, etc.[1–3] Commercially available uncooled infrared (IR) cameras using focal plane arrays based on thermal detectors are known for low detectivity, slow integration time, and small array format.[3] The need for affordable sensors with higher resolution and larger array size is challenging current IR sensor community to explore alternative material systems. Recently, several groups reported prototype *p-i-n* diode arrays using InAs/GaSb superlattices (SLs),[4,5] imaging body temperatures up to 170 K,[4] opening possibilities of their use as uncooled mid-IR sensors.

However, room temperature operation of SLs as minority carrier devices in the *p-i-n* configuration requires a sufficient quantum efficiencies (η) and differential resistances at zero bias (R_0A). Quantum efficiency is determined by the absorption of SL and the minority carrier diffusion length, while R_0A is limited by the dark current resulting mostly from Auger and defect-driven recombination processes.[1,6] For the mid-IR detection (3–5 μm), SLs have achieved reasonably good absorption ($\sim$2000 cm^{-1}) due to the intrinsically large electron-hole wave function overlap. Therefore, major objective during growth optimization process is controlling residual background carrier concentrations in unintentionally doped regions of *p-i-n* diodes to increase the minority carrier lifetimes and reduce the dark current. According to Kinch,[7] residual doping concentrations must be reduced below 10^{14} cm^{-3} in order to produce depletion region widths on the order of 5 μm, which are necessary to achieve sufficient η for room temperature operation. Recently, several papers addressed the issue of lowering the residual carrier doping concentrations,[8–10] however, have not reached the desired goal. In particular study of longer wavelength SLs, Bürkle *et al.*[10] showed the importance of growth temperature (T_g) on the intrinsic carrier type and doping concentration. Moreover, T_g controls anion exchange mechanism that affect interfacial roughness between the layers, which, in turn, determines both the majority and minority carrier transport properties along the in-plane and growth direction.[10,11] In this paper, we study the effect of molecular beam epitaxy (MBE) growth temperature on the residual carrier properties in typical mid-IR 7 ML InAs/8 ML GaSb SLs, which provides the cutoff wavelength around 4 μm ($\sim$300 meV).

The InAs/GaSb SLs were grown by MBE on unintentionally doped (residual *p*-type) GaSb (100) wafers. Followed by 0.5-μm-thick GaSb buffer layer growth, a SL consisting of 110 repeats of 7 ML InAs/8 ML GaSb layers were grown at substrate temperatures ranging from 370 to 430 °C. Details of the growth procedure were reported elsewhere.[12,13] To grow the intended layer thickness of InAs and GaSb, the growth rates of 0.9 and 0.25 Å/s were used, respectively. The V/III beam flux ratio of the InAs and GaSb was fixed at 3, which previously produced smooth surface morphologies.[13] The interface type in this study was not controlled. Intended SL structures were confirmed by high resolution x-ray rocking curve, measuring the periods and strains around (45.0 ± 0.3) Å and (-0.15 ± 0.05)%, respectively.

To determine residual carrier properties, conventional variable temperature Hall effect measurements were performed on each grown sample. Ohmic contacts were placed at the corners of roughly 1×1 cm^2 van der Pauw samples, and variable temperature (10–300 K) resistivity and single field (0.5 T) Hall measurements were performed in a guarded direct-current system. Because residual *p*-type GaSb substrates can show significant conduction throughout the temperature range (10–300 K), the Hall mobilities (μ) from SL samples were compared with those obtained from a typical GaSb substrate (Fig. 1). Above 20 K, the mobility is dominated by the parallel conduction channel through the *p*-type substrate. Below 20 K, the SL and substrate present competing parallel conduction channels as the number of mobile holes in the substrate increases. Below 10 K, holes in the substrate move only via hopping with negligible mobilities. Therefore, we take the 10 K results as the characteristic of the SL themselves.

Figure 2 plots the 10 K carrier densities as a function of T_g. The lowest carrier density of 1.8×10^{11} cm^{-2} ($\sim$3.6

[a]Electronic mail: heather.haugan.ctr@us.af.mil.
[b]Permanent address: Department of Physics, University of Dayton, Dayton, OH 45469.
[c]Permanent address: Department of Physics and Astronomy, Bowling Green State University, Bowling Green, OH 43403.

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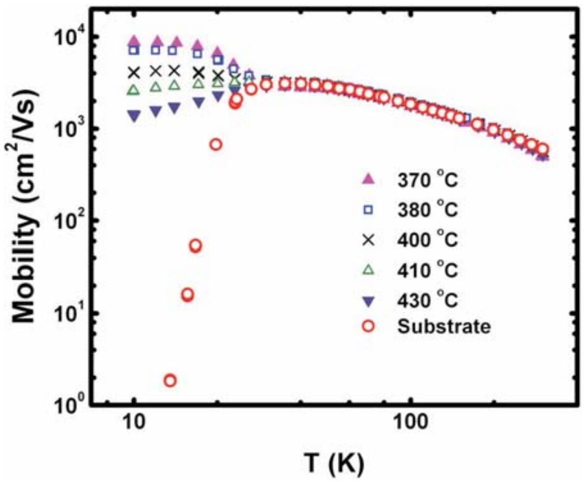

FIG. 1. (Color online) The in-plane carrier mobility as a function of growth temperature for the 7 ML InAs/8 ML GaSb superlattices (SLs) grown between 370 to 430 °C and the GaSb substrate. Below 20 K, substrate conduction is frozen out so that the 10 K mobility is dominated by the SLs.

$\times 10^{15}$ cm^{-3}) was obtained from the SL grown at 400 °C. At T_g below or above 400 °C, the carrier density slightly increases at a most a factor of 2, but still stays in the low 10^{11} cm^{-2} (10^{15} cm^{-3}). All SL samples in the series were residual p type and remained as p type throughout the entire temperature range studied (10–300 K). This differs from the results of an earlier magnetic transport study done on longer wavelength of 10 ML InAs/5 ML $In_{0.25}Ga_{0.75}Sb$ (binary/ternary) systems.[10] The SLs grown at 360 and 380 °C were residual n type with electron concentration as high as the low 10^{16} cm^{-3}, meanwhile those grown at 400, 420, and 440 °C were residual p type with hole concentration maximum as high as the high 10^{15} cm^{-3}. In our experience, InAs normally grows n type and GaSb p type, therefore the difference seen in Ref. 10 might be coming from using different ratios of the GaSb-to-InAs layer widths; for our SLs, the ratio is 8 to 7 that favors p-type SLs, whereas the ratio in Ref. 10 is 1 to 2 that favors n-type SLs as observed. The behavior seen in Fig. 2 can be explained using similar arguments claimed in Bürkle *et al.*[10] Lower T_g lead to an excess of group V elements, which leads to group III vacancies. Increased incorporation of residual group IV impurities at vacant group III sites, where they act as donors, causes the SLs to be less p type as seen in Fig. 2 from 370 to 400 °C. With increasing T_g, the excess of group V atoms decreases, the number of group III vacancies decreases, so that the residual group IV impurities begin to incorporate at group V sites, where they act as acceptors. The overall trend is to make the SLs more p type, as seen in Fig. 2 above 400 °C. However, the residual carrier density in Ref. 10 changes by about an order of magnitude, whereas the variation in Fig. 2 is only by about a factor of 2. The latter observation is consistent with the fact that our SLs are p type to begin with.

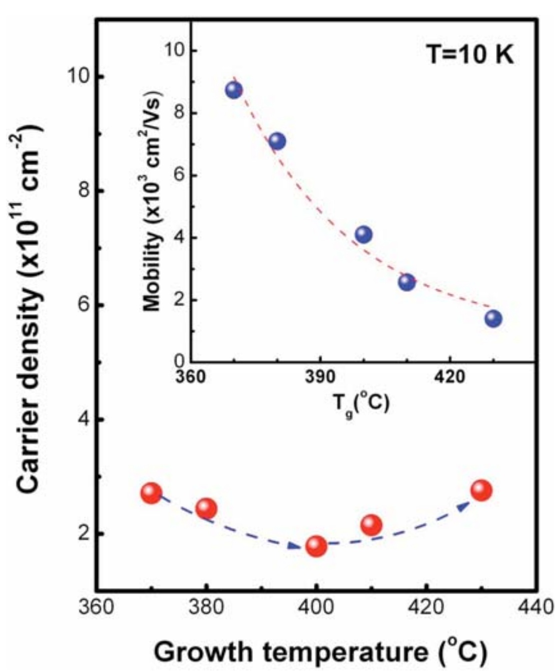

FIG. 2. (Color online) The carrier density and in-plane mobility (the inset) as a function of the growth temperature (T_g) for the 7 ML InAs/8 ML GaSb superlattice measured at 10 K.

In photovoltaic devices, efficient minority carrier conduction and diffusion require high carrier mobilities for higher speeds of operation and greater diffusion lengths. The mobilities of minority and majority carriers in the growth and in-plane directions are mainly limited by interface roughness scattering (IRS).[11] Therefore, the in-plane majority carrier mobilities are indicative of the sample quality.[11] The inset of Fig. 2 shows the 10 K in-plane majority hole mobility as a function of T_g. As the T_g increases from 370 to 430 °C, the in-plane hole mobilities decrease sixfold. In our previous mobility study of the n-type SLs with increasing InAs layer width,[11] the mobilities were IRS limited as evidenced by the temperature independence of their mobilities around 10 K and their classic sixth-power dependence on layer widths. Because hole masses in the growth direction are an order of magnitude greater than the electron mass, holes are better confined in the GaSb layers than are electrons in the InAs layers. Hence, holes should be more sensitive to interface roughness fluctuations than are the more itinerant electrons. The degree of interface roughness can be extracted based on the available models,[14,15] according to which mobility μ is given by

$$\mu = \frac{eL^6}{\pi^5 \hbar \Delta^2 \Lambda^2} \times F(\Lambda, k_F)^{-1},$$

where Δ is the interface roughness, Λ the coherence length of interface fluctuations, L the width of the current-carrying layer, k_F the Fermi wave vector, and $F(\Lambda, k_F)$ is a scattering integral that depends on screening. Using $k_F^{-1} \approx 890$ Å and the calculated screening length of $q_s^{-1} \approx 100$ Å, we find $F(\Lambda, k_F) \approx 0.010$, so that $\mu \approx (1.04 \times 10^7\ \mathrm{cm^2/V\,s})/(\delta\lambda)^2$, where λ and δ are the numerical values of Λ and Δ in angstroms. With 1 ML roughness, $\delta = 3.048$, so that $\Lambda = 12$ Å at 370 °C and $\Lambda = 28$ Å at 430 °C; for 2 ML roughness, $\delta = 6.096$, and the Λ numbers are halved. Therefore, for 1–2 ML roughness, Λ is on the order of 10–30 Å, which agrees with our estimates on n-type SLs for mid-IR detection.[16] Overall, the mobility decrease can be ascribed to the increase of either Λ or Δ.

In addition, photoluminescence (PL) measurements were used to evaluate the optical quality of the SLs and to establish a correlation between the observed electrical properties and optical quality. The PL peak corresponds to the band gap

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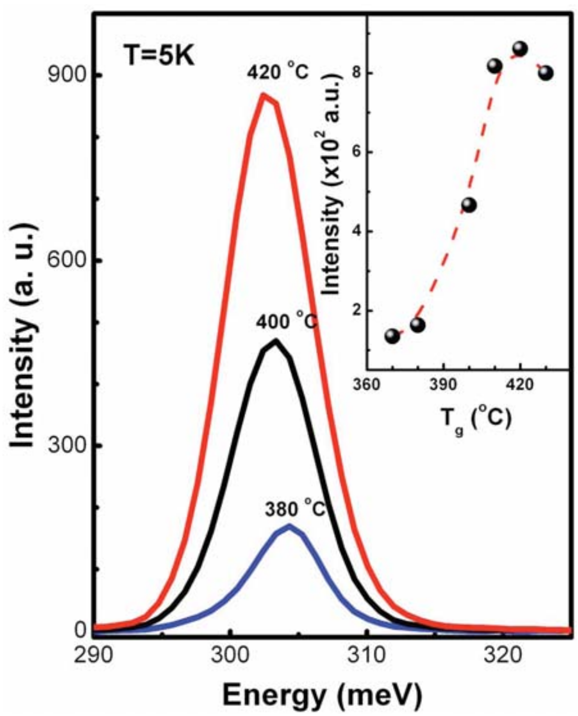

FIG. 3. (Color online) The 5 K photoluminescence spectra of three 7 ML InAs/8 ML GaSb superlattices grown at 380, 400, and 420 °C, respectively. The inset shows the PL intensity as a function of growth temperature (T_g).

recombination of photoexcited minority carriers (here, electrons) in the InAs layer with residual holes in the GaSb layer, which makes the PL intensity sensitive to the electron-hole wave function overlap as well as to the presence of any nonradiative recombination processes. The PL measurements were performed by exciting the sample with a diode pumped neodymium vanadate (Nd:YVO4) laser at 532 nm. The sample was placed in an optical Janis cryostat and the emitted PL intensity was detected with a nitrogen cooled InSb detector. Figure 3 shows three representative luminescence spectra with absolute intensities collected from the SLs grown at 380, 400, and 420 °C. For all SLs in the series, the peak energy and the full width at half maximum (FWHM) are around 300–305 meV and 6–7 meV regardless of T_g. Since the peak position remains constant, the confinement potential for the carriers is independent of the T_g, from which we infer that the interface abruptness characterized by Δ does not change. This is further corroborated by the invariance of the FWHM of the spectra in Fig. 3. By contrast, the PL spectra in the study of Bürkle *et al.*[10] show a pronounced red shift of 11 meV, which they attributed to the reduced abruptness of their SLs through interface interdiffusion involving anion interchange. Our findings suggest that the behavior of the mobility in the insert of Fig. 2 is due to the increase in the correlation length Λ rather than in the height Δ of interface fluctuations.

The inset of Fig. 3 plots the 5 K PL intensity as a function of T_g showing fourfold increases in intensity as T_g increases from 370 to 400 °C, and reaching the maximum at 410 °C and leveling off, which is similar to the behavior observed in Ref. 10. Bürkle *et al.* used different mechanisms to explain the PL behavior in their *n*- and *p*-type samples. In their *p*-type samples, the intensities saturate, which is explained by the nonradiative recombination of minority electrons that diffuse to the SL surface. Since such recombination is independent of hole concentration, the PL intensity is independent of T_g. In our case, the diffusion length $L_n=\sqrt{k_B T \mu_n \tau_n / e}$, so with the electron mobility $\mu_n = 3000\ \mathrm{cm^2/V\,s}$, $T=10$ K, and the minority carrier lifetime $\tau_n = 10^{-9}$ s,[17] we find $L_n = 5000$ Å, which is on the order of the SL thickness, consistent with the mechanism explained above. For our *p*-type samples, we explain the initial fast rise in the PL intensity by a rapid decrease in the number of nonradiative recombination centers such as vacancies in the SL itself, which agrees with the behavior of the carrier concentration data in Fig. 2. As the samples become more *p* type, the PL intensity is also enhanced by the increase in the number of holes for electrons to recombine with. Then, as nonradiative defects are reduced, surface recombination becomes the rate limiting step and the PL intensities saturate.

In summary, we studied the effect of T_g on the density and mobility of residual carriers in typical midinfrared 7 ML InAs/8 ML GaSb SLs grown by MBE. All SL samples were residual *p* type, with carrier densities in the low 10^{11} cm^{-2} over the 370–430 °C growth window studied. The minimum density of 1.8×10^{11} cm^{-2} for the SL grown at 400 °C is explained by the incorporation of residual group IV impurities first at cation vacancies and then at group V sites. With increasing T_g, the in-plane carrier mobility decreased sixfold due to increased interface roughness and most likely the increase of the correlation length. At the same time, the luminescence intensity increased due to fewer nonradiative defects at higher T_g.

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